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METALS & ALLOYS

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The Magazine of Metallurgical Engineering

INCLUDING
CURRENT METALLURGICAL ABSTRACTS

PULVERIZED COAL FIRED MALLEABLE IRON MELTING FURNACE IN PLANT OF LINK-BELT COMPANY

VOLUME 4

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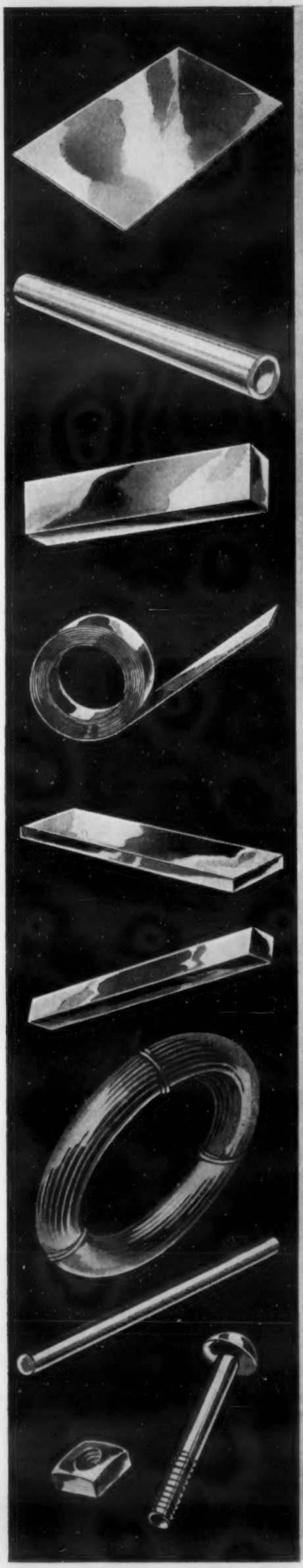
NUMBER 2



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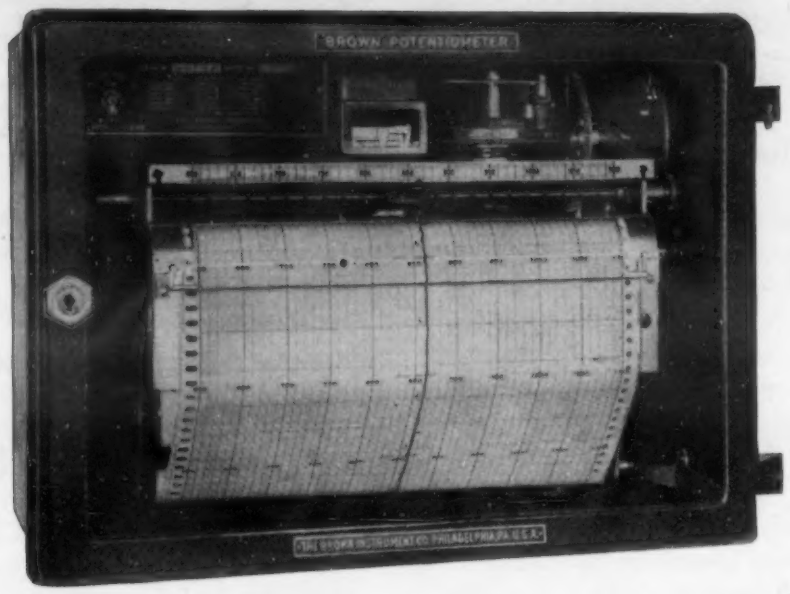

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EDITORIAL COMMENT

Ben Franklin

WE RECENTLY sat in a committee meeting that was planning a technical session on a metallurgical subject. The field is one of great commercial importance but the methods now used are rule of thumb. The practical man very much wants to have science applied to bring order out of chaos. Some very high-brow mathematical analyses have been worked out that appear to have some potential bearing, but they have chiefly been based on the text book principles of mechanics of elastic bodies and do not fit the actual case, which involves plastic deformation. Before the high-brow can add anything valuable to the subject, he must understand the actual variables well enough to make correct postulates upon which to build a mathematical theory that will be of any real use.

The plan under discussion was to get papers written that would attract attendance from both groups so as to bring the practical men in touch with the high-brows and vice versa. The committee members are well acquainted with individuals of both types. It foresaw no difficulty at all in getting the practical men not only to listen to the high-brows but also to utilize anything they might suggest that was in shape to utilize, even if the practical group itself had to dig very deeply into theory in order to grasp it. The difficulty to be overcome, to the minds of the committee, was chiefly that of getting the high-brows to pay any attention to a practical problem that required a little different angle of attack than the one they are used to.

The nonchalant way in which this committee took it as obvious that industry was quite ready and even anxious to "go high-brow" for a practical purpose as soon as it could see how to do so, surprised us for a moment. On second thought, however, we recalled that the progressive open-hearth man has become quite a physical chemist in the last five years, because he has had interpreters who have brought the high-brow and the operator together.

Our pressing need is not so much for more Einsteins and Schrödingers who develop new broad generalities without any attention to whether those generalities are capable of application, as it is for more interpreters who can see how present theoretical knowledge will fit into present practical problems, or, alternatively, show the gaps in fundamental knowledge that must be filled before the theory is complete enough and sound enough for interpretation. When such gaps are made evident, industry is very likely to arrange to get them filled. There is, we think, more real advance in pure physics, and in development of refined apparatus for getting quantitative data, being made by commercial organizations than in all the universities put together. Radio and geo-physical prospecting are cases in point.

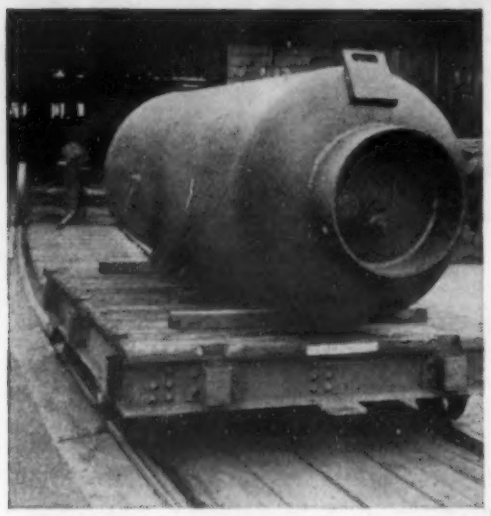
More people with the point of view of Ben Franklin would help a lot. When Franklin ran across a new scientific fact or reasoned out a scientific problem, he simultaneously saw some need which could be filled by a simple application of his knowledge. When he found that electricity would flow down his kite string, that fact was not a mere abstraction to him. He connected it with a need and the lightning rod resulted. He looked into combustion, writing a paper on "Causes and Cures of Smoky Chimneys," but his interest was not merely academic, for he put his theories to work in the Franklin stove. One of his biographers comments that he never asked for a patent on any of his inventions. He went ahead and manufactured them without delay.

He was a believer in swapping technical information. He founded the American Philosophical Society, and was a member of the French Academy of Science. Even when he was engrossed in diplomatic and other civic duties, he kept in touch with all branches of science. He was up on physics, mathematics, natural history, medicine (being especially interested in the common cold and in inoculation against smallpox), was something of an authority on earthquakes and was at least interested in chemistry, being a friend of Priestley. He tried to do something about all these things, in order to put his knowledge to some use, though it is not on record that he made any very direct use of his knowledge of earthquakes, that probably being his own personal diversion in the realm of pure science.

In his day, there was not so much science but that an alert mind could keep in touch with all of it. We imagine that were he living today, he would be an eager reader of our metallurgical abstracts, and we'll bet he'd see more ways to apply in practice the facts there brought out than most of us moderns do.

His point of view is the greatest asset a research man could have. As an editor, we wish, too, that more people could write as he could. Besides grasping the relation of theory to practice, he could (and naturally so from his early training as printer, publisher and journalist) express his ideas both clearly and concisely. His scientific articles would probably be pretty fairly high-brow but he would not fail to point out the practical bearing of the theory he was dealing with.

Ben Franklin might well be the patron saint of any research laboratory, and a model for scientific writers.—H. W. GILLETT



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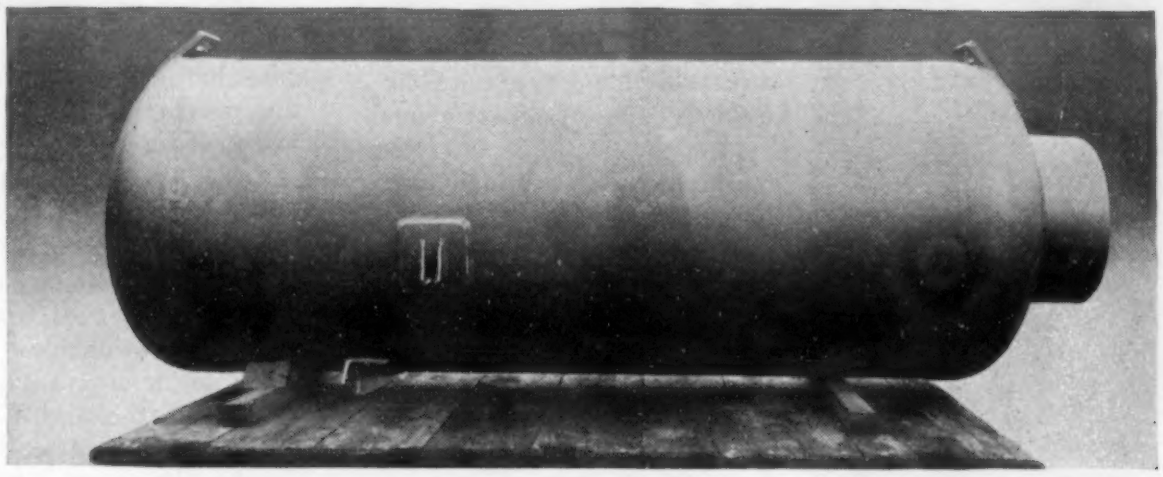

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Non-Rusting & Heat Resisting 34% Chromium Alloy Cast Irons

By E. HOUDREMONT and R. WASMUHT

R. Wasmuht

E. Houdremont

Three cast iron alloys containing about 34% chromium and varying carbon content showed excellent corrosion resistance, erosion resistance and heat resistance.

UNDER the term "non-rusting cast-iron," various alloy cast-irons are already on the market which possess a slightly improved corrosion resistance as compared with ordinary gray iron, due to the addition of small quantities of chromium, nickel, etc. Below, it is not proposed to deal with alloys of this nature, but with such as contain comparatively high percentages of chromium and are truly rust-proof, having been developed as a parallel to the well-known non-rusting steels.

It is a well-known fact that in the case of chromium-iron alloys a sudden very marked increase in corrosion resistance occurs when the chromium content reaches 13% or thereabout. It is further known that if carbon be added in gradually increasing quantity to iron-chromium alloys containing from 13 to 15% Cr, their resistance in the annealed state to corrosive attack diminishes with increasing carbon content. The reason for this may be of a twofold nature:

1. Owing to the formation of chromium carbides, chromium is abstracted from the matrix of the solid solutions. In proportion as the matrix becomes poorer in chromium, its ability to resist corrosion diminishes.
2. The embedded carbides are possibly less resistant to corrosion than the ground mass, and may in addition produce local electro-chemical action, thus starting stronger corrosive attack.

The latter cause is probably of secondary importance in the case under consideration, since chemical compounds are as a rule characterized by great stability, and the danger of electrochemical action being set up is also very slight, as no great difference in potential exists between the high-chromium carbides and the matrix, which is also rich in chromium. The decrease in corrosion-resistance of iron-chromium alloys as a consequence of increased carbon content must therefore be ascribed mainly to the first-named cause, viz: the abstraction of chromium from the matrix.

In connection with the high-chromium cast irons investigated by the authors, considerable importance would appear to be attached to the trigonal chromium carbide of Cr_7C_3. If the chromium combined with the carbon in the form of this carbide, 1% carbon would abstract about 10% chromium from the matrix. As a matter of fact, however, the separated carbides are probably not in this pure form, as iron will be substituted for part of the chromium. The investigations of A. Westgren, G. Phragmén and Tr. Negresco[1] have confirmed the solid solution nature of the chromium carbides, anticipated by P. Goerens and Stadeler[2] as early as 1907. The chromium carbide Cr_7C_3 is therefore able to absorb a limited quantity of iron by atomic substitution, without changing its crystallographic nature. In practice, therefore, the amount of chromium abstracted from the matrix is less than that indicated by the above calculation, but still remains quite considerable in the case of high carbon contents.

It follows from the foregoing that if it be desired to increase the carbon content of iron-chromium alloys

*Translated by the authors from *Krupp'sche Monatshefte*, Vol. 12, 1931, pages 331-337.

[1] *Jernkontorets Annaler*, Vol. 111, 1927, page 216.
[2] *Metallurgie*, Vol. IV, 1907, page 18.

TABLE I

Medium of attack	Test temperature °C.	Test Results Alloy a	Alloy b
Nitric Acid HNO_3			
7%, sp. gr. 1.04	20	fully resistant	fully resistant
7%, sp. gr. 1.04	boiling	fully resistant	fully resistant
37%, sp. gr. 1.23	20	fully resistant	fully resistant
37%, sp. gr. 1.23	boiling	satisfactorily resistant	satisfactorily resistant
66%, sp. gr. 1.40	20	fully resistant	fully resistant
66%, sp. gr. 1.40	boiling	satisfactorily resistant	satisfactorily resistant
conc. fuming, sp. gr. 1.52	20	satisfactorily resistant	satisfactorily resistant
conc. fuming, sp. gr. 1.52	40	non-resistant	non-resistant
conc. fuming, sp. gr. 1.52	boiling		
Mixed Acid (% by weight)			
30% conc. sulphuric acid + 5% conc. nitric acid + 65% water	50	fully resistant	fully resistant
Sulphurous Acid SO_2			
saturated aqueous solution	20	fully resistant	satisfactorily resistant
Phosphoric Acid H_3PO_4			
10%, sp. gr. 1.05	boiling	fully resistant	fully resistant
45%, sp. gr. 1.30	boiling	fully resistant	fully resistant
80%, sp. gr. 1.64	boiling	slightly resistant	slightly resistant
Acetic Acid $C_2H_4O_2$			
10%, sp. gr. 1.01	boiling	fully resistant	fully resistant
80%, sp. gr. 1.07	boiling	fully resistant	fully resistant
100%, sp. gr. 1.05	boiling	fully resistant	fully resistant
Lactic Acid $C_3H_6O_3$			
1.5%	boiling	fully resistant	fully resistant
Sodium Hydroxide (Caustic Soda) NaOH			
aqueous solu. 15%, sp. gr. 1.16	50	fully resistant	fully resistant
aqueous solu. 15%, sp. gr. 1.16	boiling	fully resistant	fully resistant
50%, sp. gr. 1.53	boiling	non-resistant	non-resistant
Ammonium Nitrate NH_4NO_3			
aqueous solution, 50%	boiling	fully resistant	fully resistant
Ammonium Sulphate $(NH_4)_2SO_4$			
aqueous solution, 50%	boiling	fully resistant	fully resistant
Calcium Bisulphite Ca $(HSO_3)_2$			
aqueous solution, sat. cold,			
at 12-15 atm. pressure	about 150	non-resistant	non-resistant
at 4-6 atm. pressure	about 90	fully resistant	fully resistant
Calcium Hypochlorite Ca $(OCl)_2$			
aqueous solution, saturated cold	40	satisfactorily resistant	satisfactorily resistant
Ferric Chloride $FeCl_3$			
aqueous solution, 30%	20	non-resistant	non-resistant
Potassium Nitrate KNO_3 *(Salpeter)*			
aqueous solution, 25%	boiling	fully resistant	fully resistant
Sodium Sulphide $Na_2S + 9H_2O$			
aqueous solution, 50%	boiling	fully resistant	fully resistant
Zinc Chloride $ZnCl_2$			
aqueous solution, sp. gr. 1.20	boiling	fully resistant	fully resistant
aqueous solution, sp. gr. 1.90	boiling	non-resistant	non-resistant

The terms employed in the above Table to denote various degrees of resistance correspond to the following losses in weight:

Loss in weight in g./m²/hr.	Degree of resistance
Less than 0.1	fully resistant
0.1 to 1.0	satisfactorily resistant
1.0 to 3.0	fairly resistant
3.0 to 10.0	slightly resistant
over 10.0	non-resistant

without reducing their corrosion resistance, the loss of chromium from the matrix due to the formation of carbides must be made good, in other words, the chromium content of the alloy must be increased accordingly. Now, iron-chromium alloys containing a high percentage of carbon are of particular interest, inasmuch as, with increasing carbon content, they assume the characteristics of cast iron and show better casting properties than the well-known non-rusting steels. A number of casting alloys, containing an increased percentage of chromium in conjunction with a high carbon content, have therefore been investigated by the present authors in regard to corro-

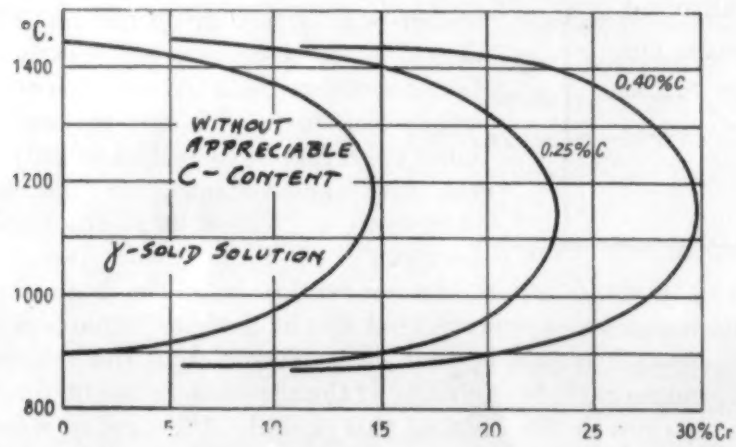


Fig. 1. Influence of carbon content on size of γ-region in iron-chromium alloys (E. C. Bain.)

sion resistance, structure and physical properties. The chromium content of these alloys was varied between 30 and 40%, with carbon between 1 and 4%.

As is well known, pure and low-carbon chromium-iron alloys containing 30% Cr are located in the purely ferritic region of the chromium-iron diagram, since a chromium content as low as 15% already suffices to cause the γ-change to disappear entirely. Fig. 1, taken from a paper by Bain,[3] shows the enlargement of the γ-region due to an increase in carbon content. Nevertheless, it is to be expected that alloys containing more than 30% Cr will consist of ferritic solid solutions, in which carbides are embedded. With regard, on the other hand, to the changes produced in the iron-carbon diagram by the addition of chromium, it has been shown by a large number of investigators, as, for instance, J. G. H. Monypenny,[4] Oberhoffer and Daeves,[5] Austin,[6] and, more recently, M. A. Grossmann,[7] that all the lines of the diagram, particularly the line of carbide separation in both the solid and the liquid solution, are shifted towards the

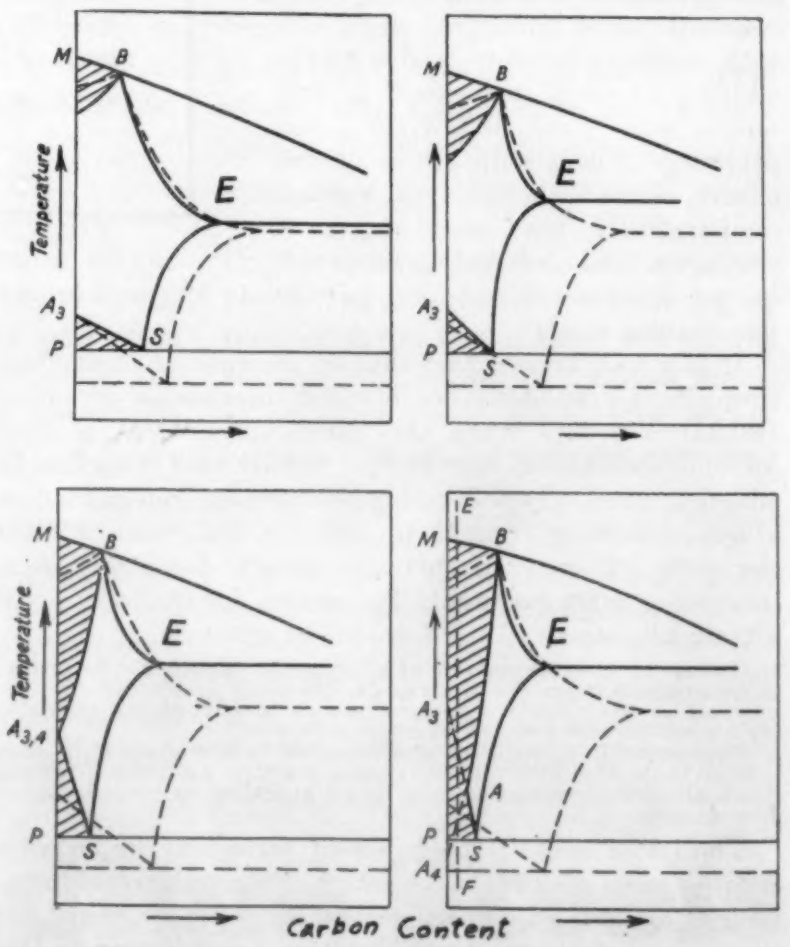


Fig. 2. Sections through the ternary iron-chromium-carbon system, corresponding to 5, 10, 14 and 18 % chromium content. Dotted lines represent the binary iron-carbon system. (M. A. Grossman.)

[3] *Transactions American Society for Steel Treating*, Vol. 9, 1926, page 9.
[4] *Journal Iron & Steel Institute*, Vol. 101, 1920, page 493.
[5] *Stahl und Eisen*, Vol. 40, 1920, page 1515.
[6] *Journal Iron & Steel Institute*, Vol. 108, 1923, II, page 235.
[7] *Transactions American Institute Mining & Metallurgical Engineers*, Vol. 75, 1927, page 214.

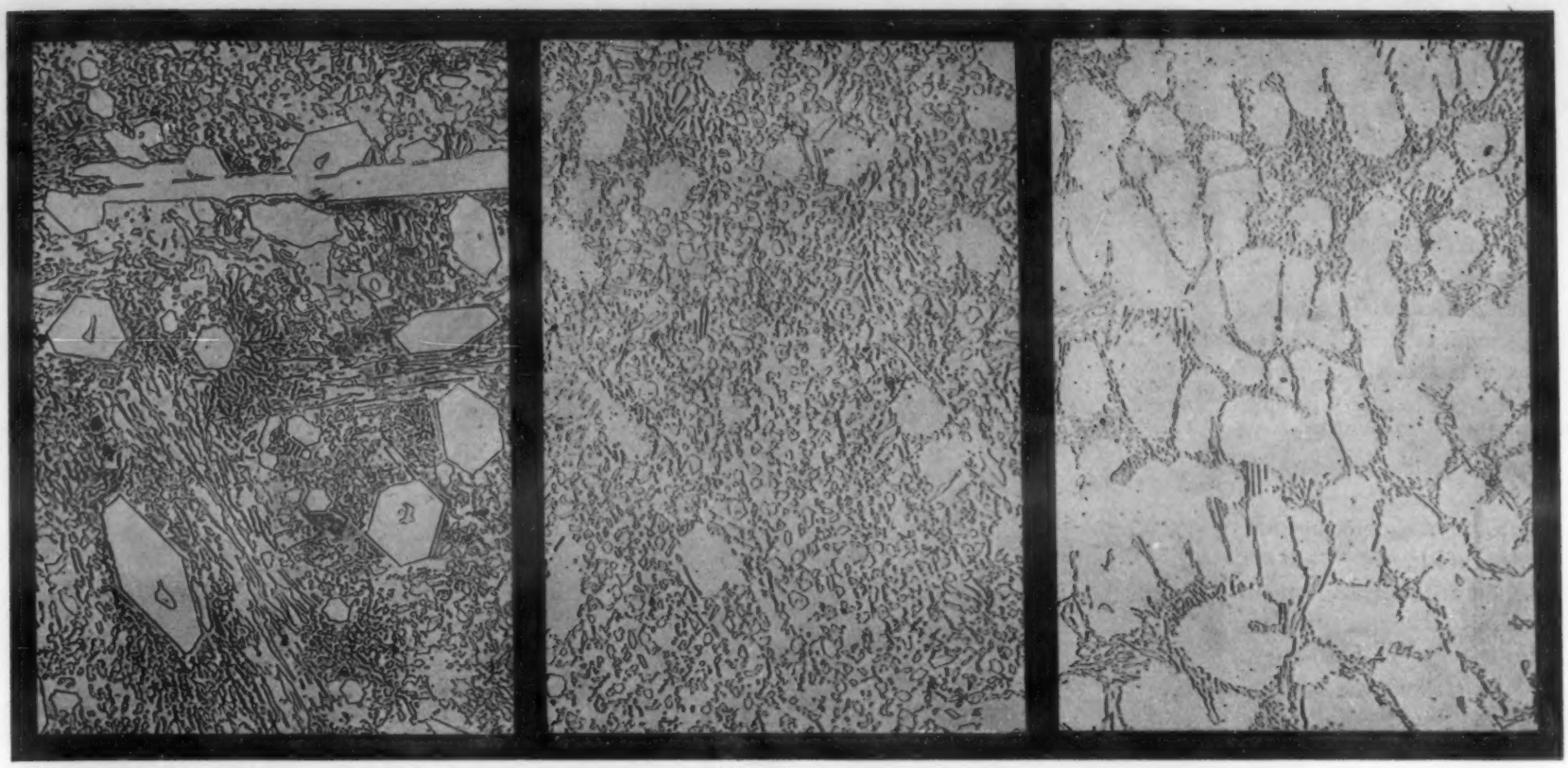


Fig. 3. Alloy c. Hyper-eutectoid. Primary cementite + eutectic. Mag. 200x.

Fig. 4. Alloy b. Approximately eutectic. Eutectic + a little solid solution (E). Mag. 200x.

Fig. 5. Alloy a. Hypo-eutectic. Solid solution (E) + a little eutectic. Mag. 200x.

left with the result that the eutectoid pearlite and the eutectic ledeburite are transferred into the region of low carbon content, as shown by Fig. 2, which is reproduced from Grossmann's paper. Alloys containing a considerable amount of chromium thus become cast irons at a low carbon content, that is to say, they are situated to the right of point E in the iron-carbon diagram.

Of the great number of alloys investigated by the authors, it is proposed to consider more closely 3 alloys, containing about 34% Cr, with varying carbon content, which offer particular interest owing to their excellent properties. The analysis of these alloys is as follows:

	C	*Si*	*Mn*	*Cr*
Alloy a	1.1%	1.3%	0.42%	33.6%
Alloy b	2.3%	1.4%	0.40%	34.2%
Alloy c	3.1%	1.2%	0.38%	34.9%

Fig. 3 shows a polished specimen of Alloy c. In a matrix consisting of a eutectic corresponding to the ledeburite eutectic are embedded primary crystals, presumably of the chromium-iron cementite $(Cr\text{-}Fe)_7C_3$ referred to earlier in this paper. The material is therefore hyper-eutectic. Fig. 4 shows the structure of Alloy b, consisting almost exclusively of the eutectic, with a few primary solid solutions (E). The material is therefore nearly eutectic, or slightly hypo-eutectic. It is thus proved that the ledeburite point, which in the case of iron-carbon alloys lies at 4.2% C, has been shifted to about 2.5% C by the addition of about 34% Cr. This is in harmony with Austin's[6] results, who found the eutectic point to lie at 2.4% C in the case of chromium contents of from 30 to 50%. Fig. 5 illustrates the structure of Alloy a. Only a little of the eutectic remains, whereas the primary solid solution has increased in quantity considerably. The material is distinctly hypo-eutectic. The position occupied by the 3 alloys within the equilibrium diagram is indicated by Fig. 6, which represents a section through the iron-chromium-carbon diagram at a point corresponding to a chromium content of about 34%. The figure shows that the 3 alloys must all possess the characteristics of cast iron. Alloy a is a hypo-eutectic, Alloy b is a eutectic, and Alloy c is a hyper-eutectic cast-iron. Steel, or cast-steel, characteristics can only be expected to result from much lower carbon contents. In spite of the considerable difference in the carbon content of the 3 alloys under consideration, their melting temperature does not differ much, being lowest (about 1278° C.) in the case of Alloy b containing about 2% C. It was further found that an addition of silicon also shifts the lines shown in Fig. 6 towards the left, whereas an addition of elements that cause the formation of austenite, such as manganese or nickel, shifts the lines in opposite direction, i. e. towards the right (Fig. 6).

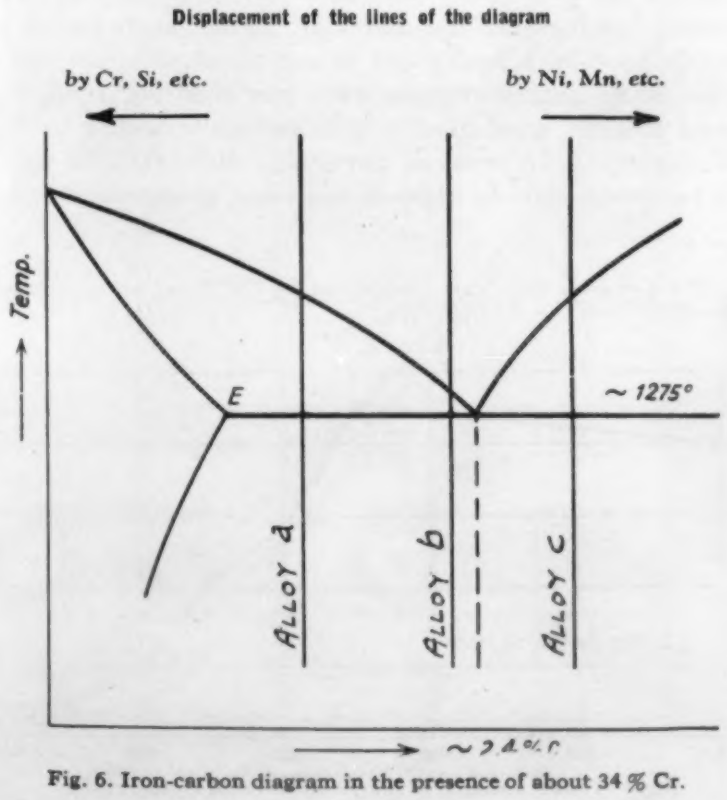
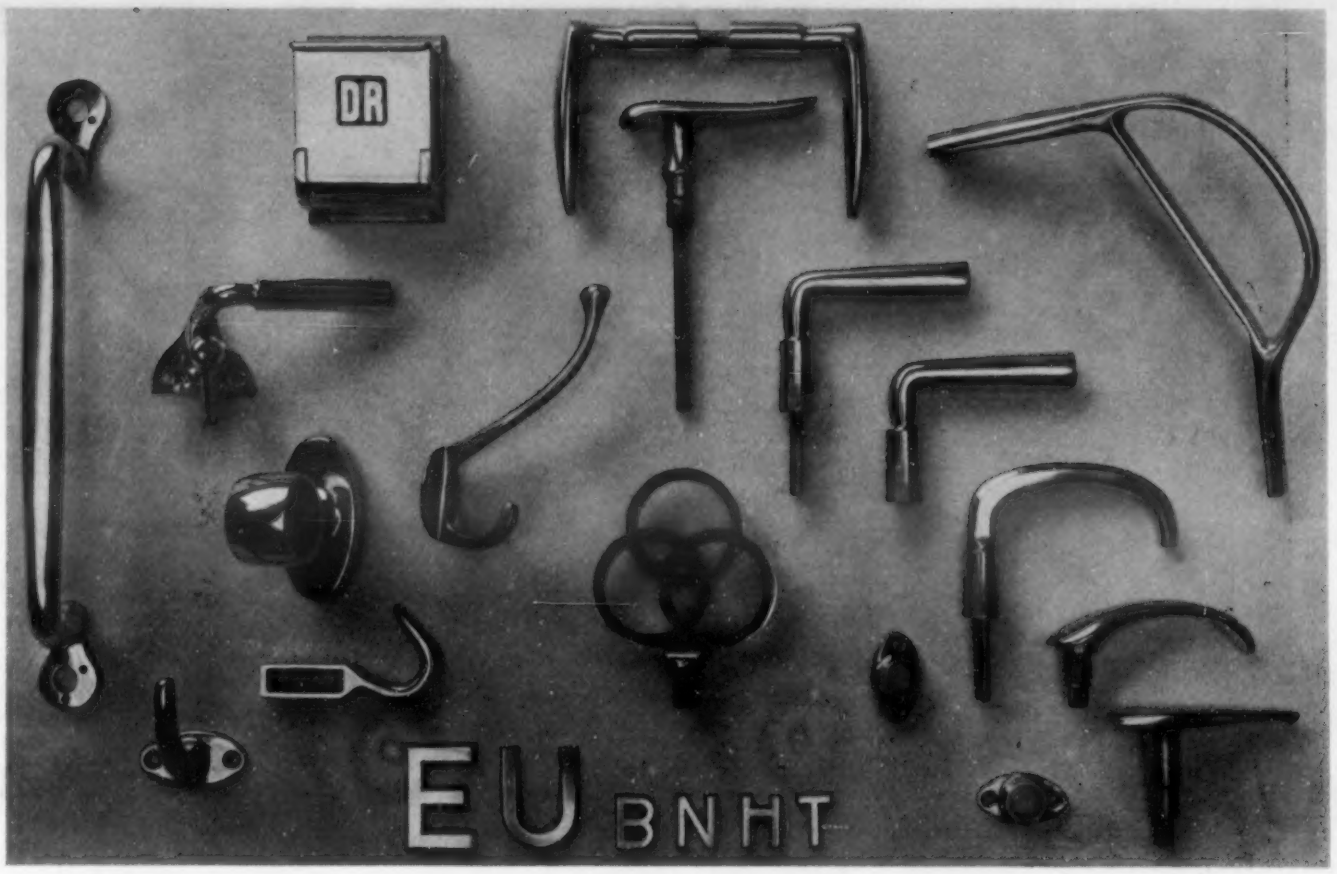


Fig. 6. Iron-carbon diagram in the presence of about 34 % Cr.

The 3 alloys under review were first tested for corrosion resistance. It was found that Alloy c showed the least resistance to atmospheric corrosion, the rough surface of the casting developing a marked rust film. In the case of Alloy a, the skin of the casting proved to be corrosion resistant, and in this respect Alloy b was found to

be only slightly inferior to Alloy a. In the polished state, all 3 alloys were perfectly resistant. Alloys a and b were subsequently tested for resistance to various chemical reagents.** The results of some of these researches are given in Table 1. Both alloys showed good resistance to attack by highly corrosive media, such as nitric acid, mixed acid, phosphoric acid, acetic acid, etc.

In addition to corrosion resistance, both alloys were shown to possess considerable resistance to scaling, being immune from oxidation up to temperatures as high as 1100° C. Coming now to the physical properties of the alloys investigated, their mechanical values are set forth in Table 2. It will be noted that in all 3 cases the tensile strength is comparatively high, considering that these alloys partake of the characteristics of cast iron. The values determined are approximately twice as high as for ordinary gray iron. The hardness increases in a marked manner with increasing carbon content. Like hard cast iron, the alloys show no elongation to speak of. Hot tensile tests were carried out with Alloy b. The results are given in Table 3 and Fig. 7. Even at high temperatures no elongation worth mentioning was noted. A noteworthy feature is the increase in tensile strength at temperatures up to about 500° C. A test piece broken at a temperature only slightly above 100° C. shows an essentially finer and tougher grain.



Polished parts made of "Nirostaguss."

TABLE II. Tensile Properties of Alloys a, b and c

	Transverse* Strength lbs./in.²	Deflection in.	Tensile Strength lbs./in.²	Brinell Hardness No.
Alloy c	74,800	.23	66,000	350
Alloy b	77,400	.23	61,500	300
Alloy a	73,000	.20	55,500	250

*Size of bar 27.6 in. long, 1.18 in. diameter. Span 23.6 in.

TABLE III. Tensile Strength of Alloy b at elevated temperatures

Temperature °C.	Tensile Strength lbs./in.²
20	58,500
500	65,500
600	40,500
800	15,500
900	9,400
1000	6,400

"Nirostaguss" as a material for objects of art.

The thermal expansion was also measured for alloys a and b. As appears from Table 4 and Fig. 8, the values found are approximately the same as for cast iron, thus being lower than for steel. As the material undergoes no transformation, it shows, contrary to mild steel and cast iron, perfectly regular expansion, so that no deformation occurs in articles subjected to even frequent changes of temperature. This property is a valuable asset in the case of applications calling for durability at elevated temperatures.

The amount of shrinkage in casting varies from 1.6 to 1.8%.

The specific gravity is 7.42 for Alloy a, and 7.43 for Alloy b.

As already stated, the alloys under consideration undergo no transformation, their matrix retaining a ferritic structure. This renders heat treatment of any kind unnecessary, except in so far as annealing is required for the relief of internal stress in the castings.

The machining of these alloys can follow the usual practice for gray iron, but takes more time, owing to the greater hardness of the material. It has been found advisable to take a heavy cut at comparatively low speed. When some practical experience has been obtained, high speed steel of good quality will generally suffice to meet all exigencies. In cases of particular difficulty, the use of the tungsten carbide type of tools has given satisfaction.

**The corrosion tests were carried out by Dr. Fritz, whose assistance the authors desire to acknowledge.

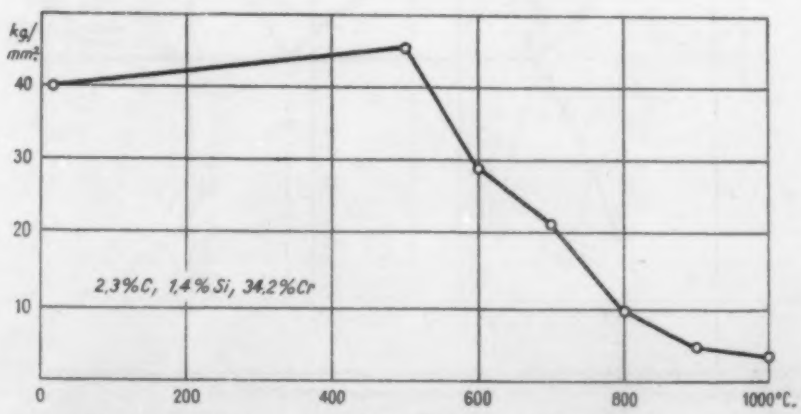
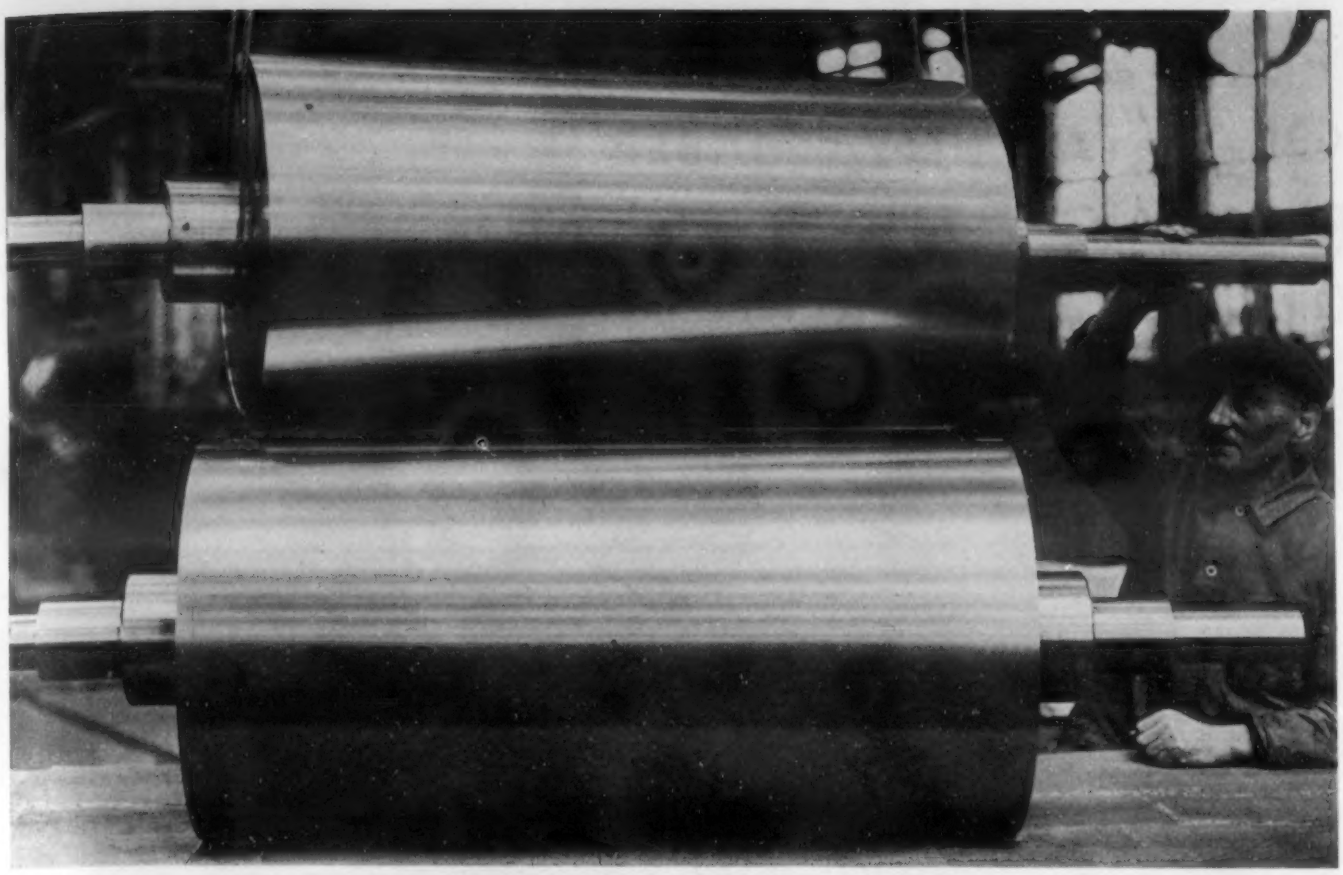


Fig. 7. Tensile strength of Alloy b at elevated temperatures.



Highly polished Hollow Rolls made of "Nirostaguss."

Electric and oxy-acetylene welding can also be carried out without any difficulty to speak of, but care must be taken in heating and cooling the work to prevent cracks starting in the vicinity of the weld. Provided the same alloy is used as a filling material, the welded joint is just as resistant to corrosion as the rest of the work, without requiring any subsequent heat treatment.

By means of a special treatment it is further possible to impart to these alloys excellent abrasion-resisting properties. They can then be used to advantage for applications calling for considerable wear resistance in addition to corrosion resistance. When so treated, however, the material is difficult to machine, the use of the tungsten carbide type of tools being indispensable.

In conclusion, a few examples of applications of these new alloys may be cited. As, owing to the high carbon content, the casting properties are relatively good, the new alloys, of a composition similar to that described above, which have meanwhile been placed on the market under the proprietary name of "Nirostaguss," lend themselves also to the production of intricate and small castings. When polished, they possess a handsome dull finish, resembling that shown by the chromium alloys, being thus particularly suitable for high-class fittings for the building trade, as well as for cocks, valves, etc. The advantage of such fittings is that they are untarnishable and so require but little cleaning. As the new alloys are further immune from attack by sea water, they are also well adapted for use in ship fittings of every kind, a field of use where a material that permanently retains its appearance without requiring much care was hitherto non-existent. Artistic castings have also been made from these alloys, which should find acceptance chiefly for articles such as fountain statues, gargoyles, etc. The fact that in addition to immunity from atmospheric attack the new alloys possess a high degree of resistance to chemical agents, as well as to scaling at elevated temperatures, opens up many other important fields of application, as, for instance, in the chemical industries, where they are used for pipes, valves, acid pumps, etc. They can further be advantageously employed for all purposes where high resistance to scaling is a requirement, being particularly suitable for the heating plates of electric kitchen ranges, etc. Larger articles, such as highly polished drying cylinders for the paper-making industry, highly polished hollow rolls, grates, etc. have also been successfully made.

Table IV. Coefficient of Thermal Expansion

Temperature °C.	Alloy a Heat	Alloy a Cool	Temperature °C.	Alloy b Heat	Alloy b Cool
	× 10⁻⁶			× 10⁻⁶	
0-100	9.8	8.7	0.100	9.4	10.0
0-200	10.2	9.9	0-200	10.3	10.6
0-300	10.6	10.5	0-300	10.8	11.1
0-400	10.9	10.9	0-400	11.3	11.3
0-500	11.2	11.1	0-500	11.4	11.6
0-600	11.4	11.3	0-600	11.5	11.8
0-700	11.6	11.7	0-700	11.9	12.2
0-800	12.2	12.4	0-800	12.3	12.9
0-900	12.7	12.8	0-900	12.7	13.6
0-1000	13.1	13.4	0-1000	13.0	14.2

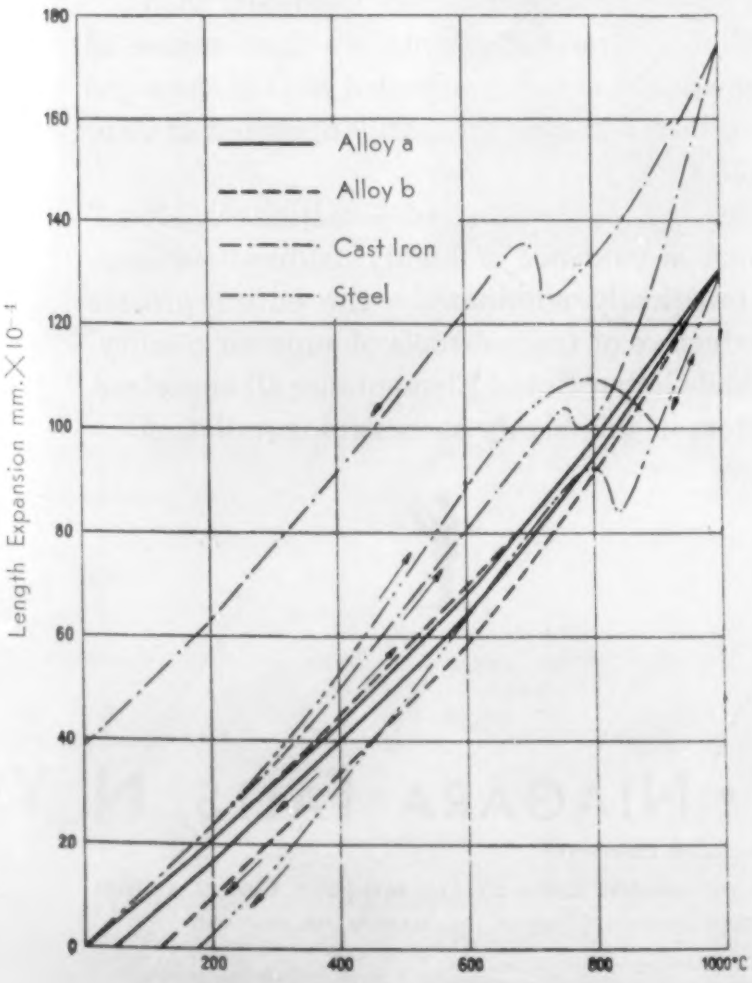


Fig. 8. Thermal expansion of Alloys a and b.

Drain Cock made of "Nirosta guss."

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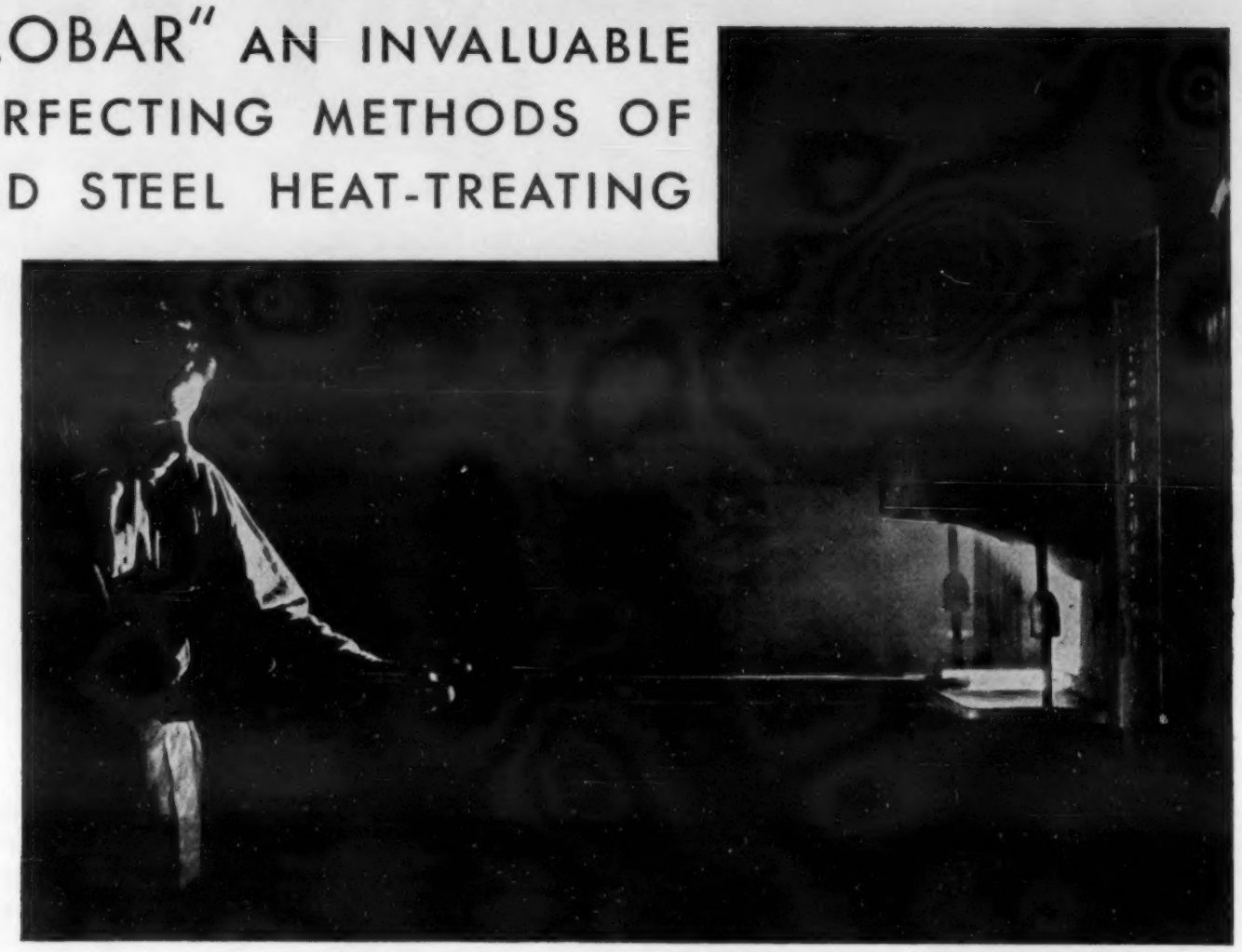

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The Formation of Localized Slip Layers in Metals

By C. W. MacGREGOR*

C. W. MacGregor

The results of a number of tests are described which were carried out in order to discuss more fully certain phenomena observed in plastically deformed metals which were reported at an earlier date.[1] Particularly, the effect of previous cold work on the formation of flow layers in aluminum was investigated. It is shown that the formation of flow layers for both hard and soft aluminum is quite different in that they become much more pronounced in the previously cold-worked material. By a certain design of compression test pieces, it is also shown how plastic flow can be localized in two well-defined layers in a non-ferrous metal, and in this way produce effects similar to those known in mild steel. The magnitude of unit strain along a flow layer in mild steel is represented by a curve.

I. Introduction

MANY POINTS of considerable interest to engineers concerning the yielding process in various materials exist. The formation of flow layers in different substances when subjected to stresses above the yield point seems to be affected by many conditions, both of testing procedure and of material treatment. When a bar of mild steel is stressed beyond the upper yield stress, flow layers begin to form along the cross-section of the test piece. Mild steel, however, is not the only material exhibiting flow layers. Under certain conditions, they may also be produced in hard copper, hard aluminum and other substances. In mild steel, one of the most important factors determining whether flow layers appear or not, is the sharpness of transition from the elastic to the plastic state as exhibited by the upper yield point in the stress-strain diagram. In a metal such as aluminum which does not have a sharp peak in the diagram, the effect of previous cold deformation seems to influence considerably the resulting slip layers after plastic deformation has once started. It is therefore the intention here to discuss first the effect of previous cold-work on the clearness of flow layers in aluminum, and then to describe the magnitude of unit strain to be expected in a flow layer of mild steel.

II. Flow Layers in Non-ferrous Metals

Compression tests were made on aluminum in order to bring out the effect of cold work described above. A number of cylindrical compression test pieces were cut from a 10 foot bar of dead-soft annealed commercially pure aluminum furnished by the Aluminum Company of America, New Kensington. These test pieces were not tested in the ordinary way between flat compression plates; but use was made of the so-called "cone compression test" originally suggested by E. Siebel and A. Pomp.[2] Following this method, the test pieces were provided with conical indentations on their ends fitting a pair of conical compression plates. This type of compression plate was used in

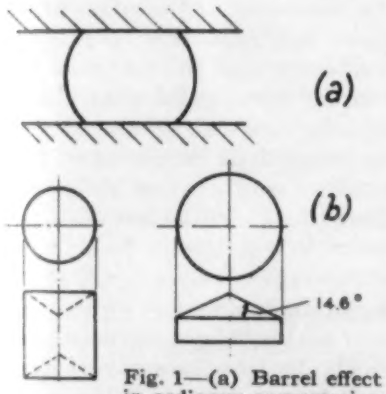


Fig. 1—(a) Barrel effect in ordinary compression tests. (b) Test piece and cone used.

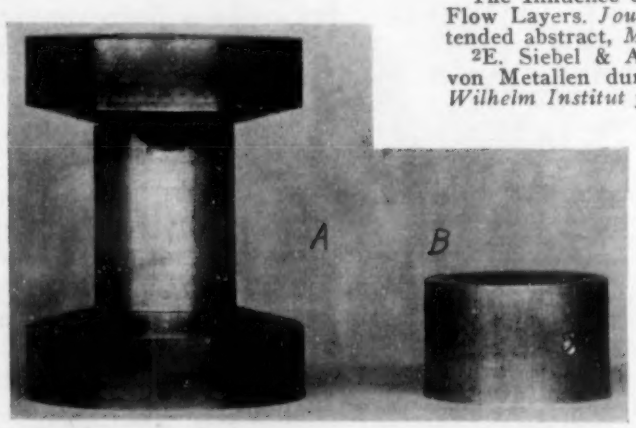


Fig. 2—Test piece in position between conical compression plates: (a) before deformation, (b) cylinder compressed 37½%.

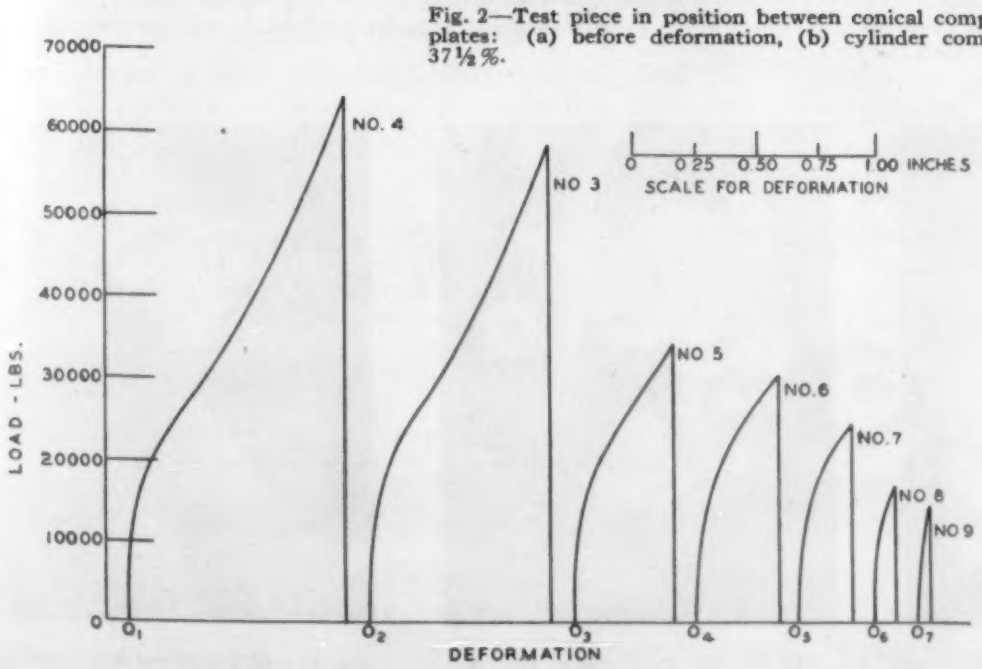


Fig. 3—Load-deformation diagram for Al cylinders tested in compression.

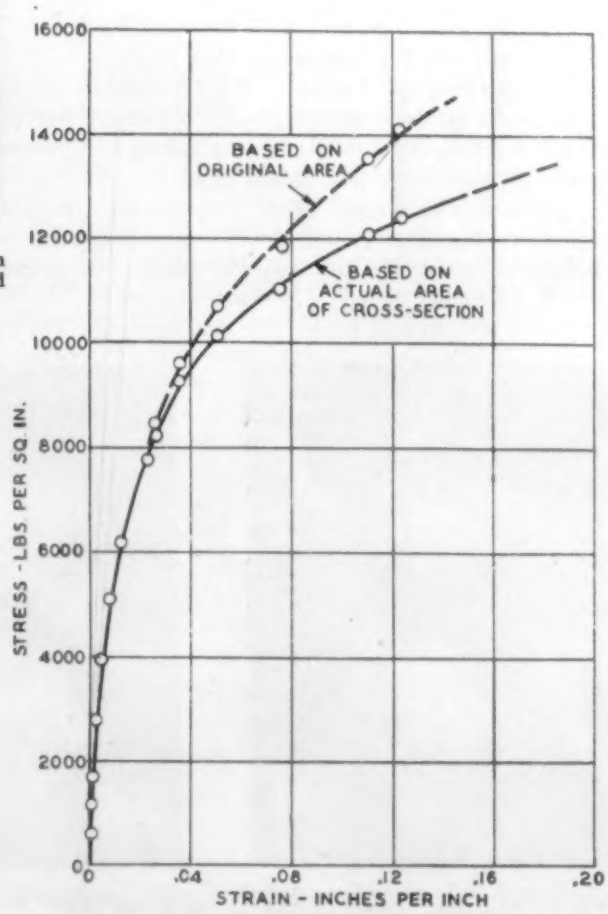
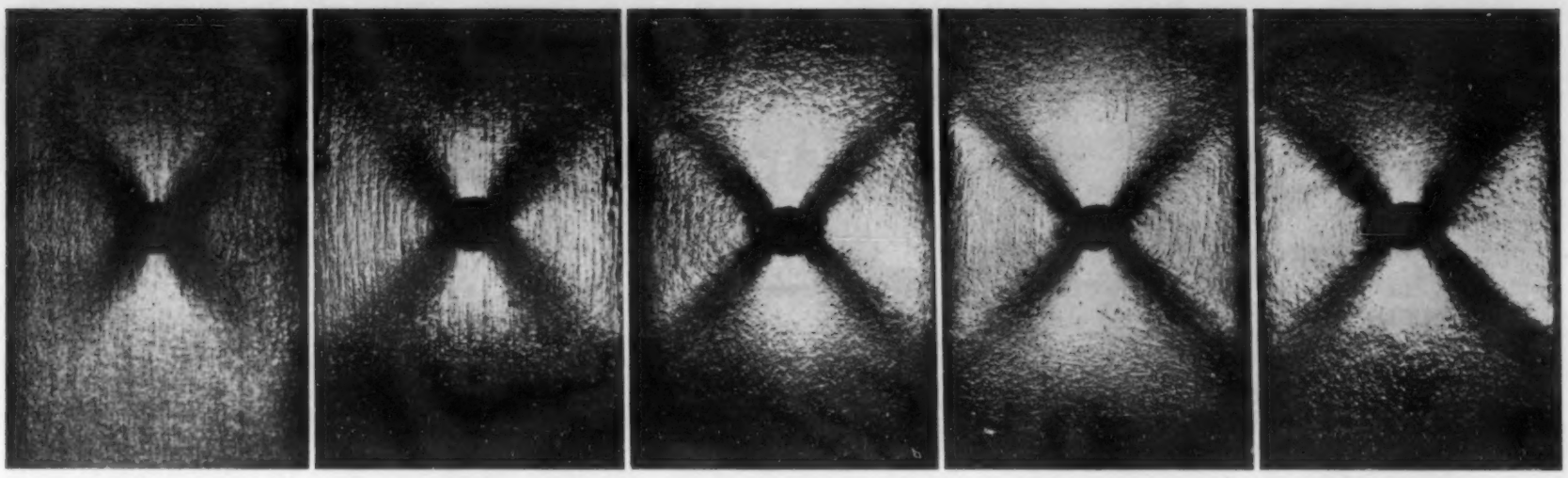


Fig. 4—Stress-strain curve for cylinder No. 5 tested in compression, using conical compression plates.

*Research Laboratories, Westinghouse Elec. & Mfg. Co., East Pittsburgh, Pa.

[1] Yield Point of Mild Steel. *Transactions American Society of Mechanical Engineers*, Vol. 53, 1931, Applied Mechanics, pages 187-200.
The Influence of Nitrogen in Mild Steel on the Ability of Developing Flow Layers. *Journal of Rheology*, Vol. 3, Jan. 1932, pages 37-52. Extended abstract, *Metals & Alloys*, Vol. 3, May 1932, page 127.

[2] E. Siebel & A. Pomp. Die Ermittlung der Formänderungsfestigkeit von Metallen durch den Stauchversuch. *Mitteilungen aus dem Kaiser Wilhelm Institut für Eisenforschung*, Düsseldorf, Vol. 9, 1927, page 157.



No. 10 0% No. 9 3.0% No. 2 26.5% No. 3 37.5% No. 4 43.8%

Amounts of pre-compression in percent

Fig. 5—Al compression test pieces. Each is compressed to 1.76% deformation after being pre-compressed to the above amounts.

order to eliminate the well known "barrel effect" shown in Fig. 1a usually received when compressing ductile metals.

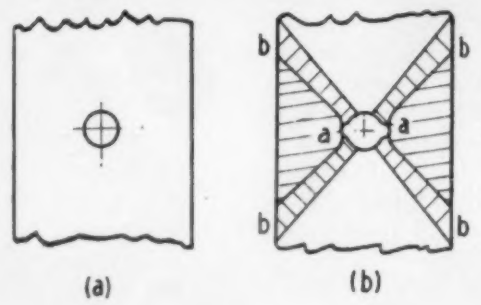
According to this method of compression, the angle of inclination of the cone to the horizontal is made equal to the friction angle for the particular material used as shown in Fig. 1b. For aluminum, this angle is about 14.6°. The component of the compressive force then neutralizes the friction force on the surface between the test piece and the compression plate, providing approximately a state of pure compression in the aluminum cylinder.

The result of this method of compression on one of the cylinders is shown in Fig. 2. It is thus seen that comparatively straight cylinders are produced after considerable deformation. This method of compression was necessary in order to provide test pieces with about the same amount of cold work throughout, since the "barrel" effect produces large variations in the amount of cold work in the same test piece.

The relative amounts of deformation for each cylinder following this method are tabulated in Table I. Fig. 3 shows the load-deformation diagrams obtained for each cylinder. A typical stress-strain diagram as taken with the use of dial gages on cylinder No. 5 is illustrated in Fig. 4.

Rectangular compression test pieces containing heads were then machined from each pre-compressed cylinder. These test pieces contained a hole in the center to provide a region of high stress concentration as shown in Fig. 7a. The sides of each rectangular test piece were highly polished. Each bar was then compressed to the same amount of 1.76% deformation, and photographed by the Schlieren method. Fig. 5 shows the results of this procedure as applied to test pieces Nos. 2, 3, 4, 9 and 10.

Fig. 6—(a) Test piece before compression. (b) Hard Al test piece after compression, showing deformed area.



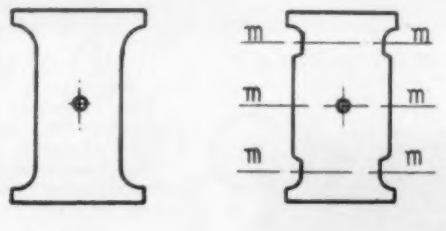


Fig. 7—(a) Test piece used on previous experiments. (b) Re-designed test piece with equal areas at sections m-m.

III. The Mechanism of Flow Around a Hole

In each case the flow layers began at points a-a in Fig. 6b and gradually curved into two directions inclined at 45° to the bar axis, thus proceeding toward points b-b. This figure also shows that at points a-a, two small semi-circular indentations appeared at the edges of the hole produced by these slip layers. The shape of the originally circular hole was deformed into the irregular shaped hole shown above. In contrast to mild steel in no case was the plastic deformation confined only to two single layers inclined in the directions of 45° to the bar axis, but some plastic deformation took place throughout the entire wedge b-a-b shown in Fig. 6b. An explanation of this difference is that for mild steel, as soon as the upper yield point is reached, the material yields quickly in thin layers extending from a to b, but since the stress is then reduced to the lower yield point, the portions of the test piece adjacent to these layers are not stressed sufficiently to become plastic. While for aluminum, even though it is cold-worked it will never have a

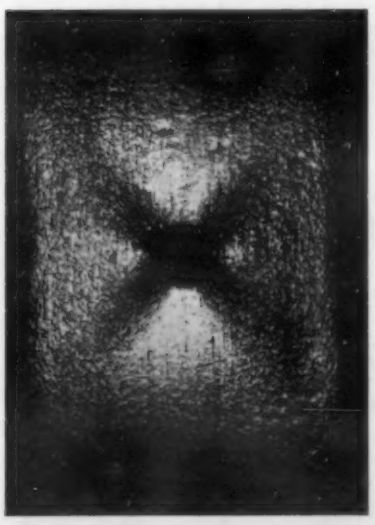


Fig. 8—Al Test Piece. Reannealed for 1 hr. at 500°C and then compressed to 1.76%.

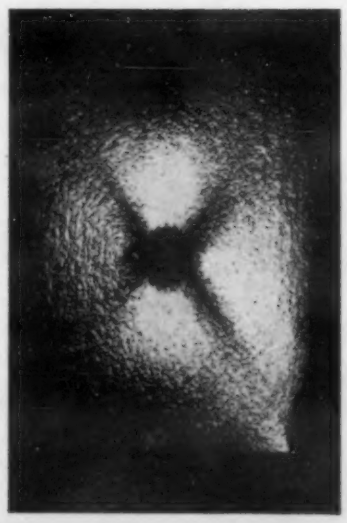


Fig. 9—Al Test Piece No. 5 prestressed to 20% deformation and then compressed to 0.61% as shown here.

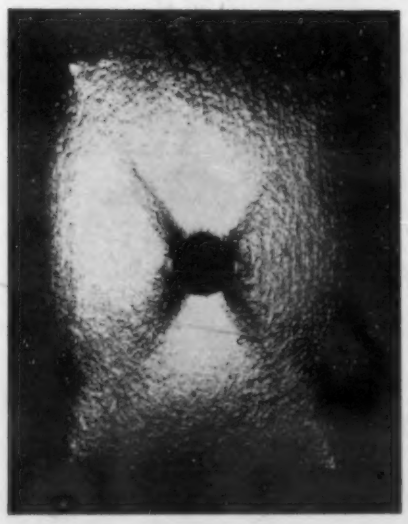


Fig. 10—Al Test Piece No. 6. Pre-compressed to 15% then to 0.99% as shown here.

Fig. 11—Al Test Piece No. 5 compressed to 4.2%.

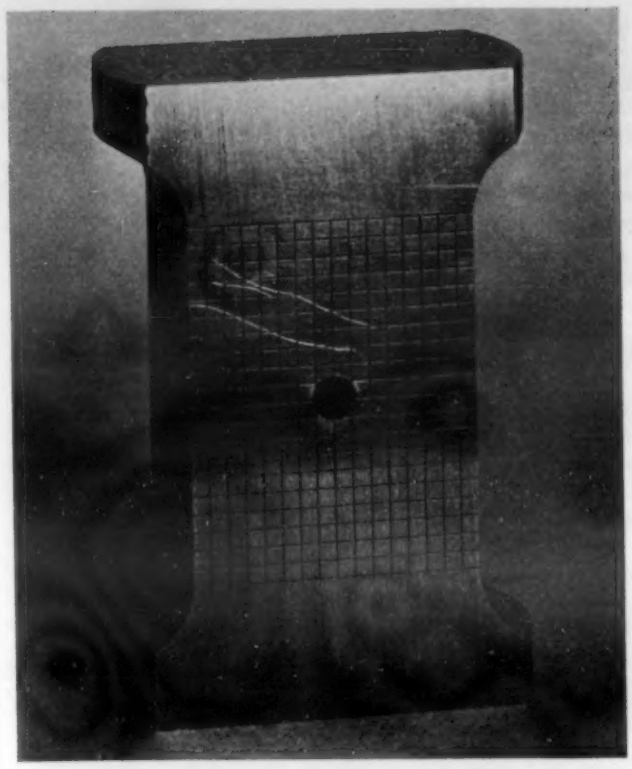


Fig. 12—Compression test piece of mild steel with ruled grid.

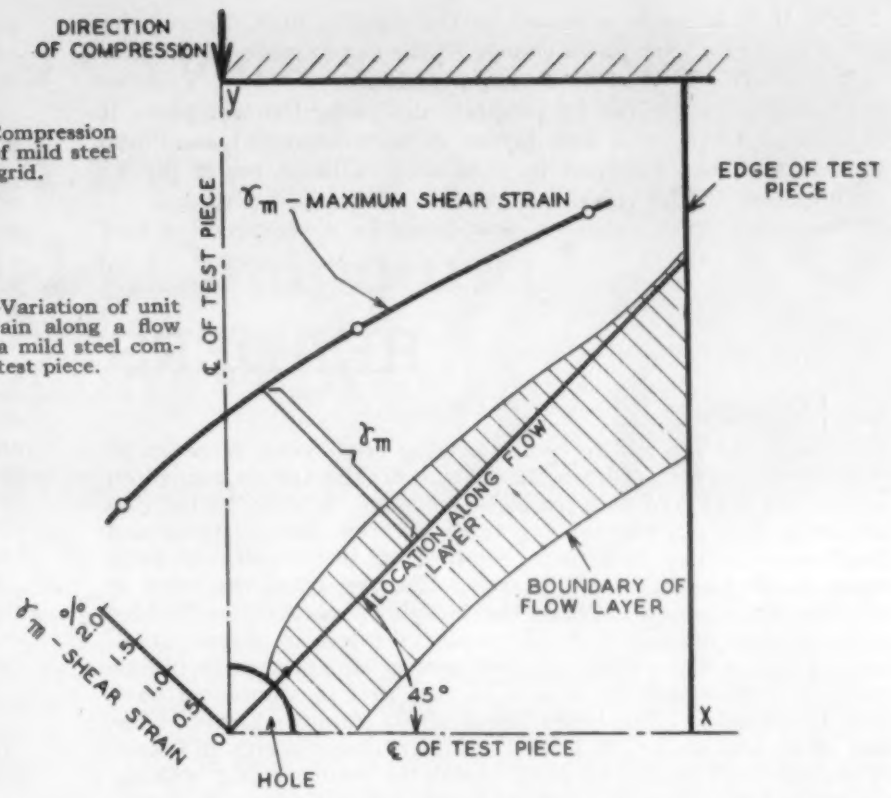


Fig. 13—Variation of unit shear strain along a flow layer in a mild steel compression test piece.

pronounced upper yield point, and as the load increases the stress at which the material yields in one portion is very soon reached in the adjacent parts of the test piece causing a more uniform yielding in the entire wedge b-a-b.

It can readily be seen from the photographs of Fig. 5, that the formation of slip layers in the various test pieces with different amounts of pre-compression was quite different. With the soft aluminum test pieces, as for example, No. 10, the formation appeared under very low loads and gradually spread over a considerable area with no clear demarkation between the elastic and plastic portions. The flow layers in this case did not penetrate far into the test piece after leaving the edge of the hole. On the other hand, the test pieces containing large amounts of pre-compression showed very clear flow layers with clear demarkation between the elastic and plastic portions. In their case the slip layers did not disappear after proceeding a short distance from the hole, but extended to the edge of the test piece. Instead of forming early in the test, they appeared, similarly as mild steel, at a certain instant, and spread quickly over the bar. It can be seen from the photographs that the effect of the stress concentration due to the hole is much more far-reaching in the harder aluminum than in the softer material.

This illustrates that the effect of a hole in a work-hardened or pre-compressed aluminum is to localize a subsequent plastic deformation in a comparatively thin layer, while in a soft aluminum the plastic deformation is distributed over a larger area around the hole. A hole may become more dangerous in work-hardened metal, because on account of the large shears in the thin layers, shearing cracks may be produced.

In addition to the tests already described, two test pieces having a specially designed shape shown in Fig. 7b were cut from work-hardened aluminum (prestrained by 20 and 15% respectively). These test pieces had the same cross-sectional area in the plane of the hole as at some distance from it, so that the *average* stress in the cross-section containing the hole was the same as in some distance from it. This was provided to eliminate an additional increase of the average shearing stress in the actually formed slip layer due to the presence of the hole. To have a good comparison with the influence of a hole in a very soft metal, in Fig. 8 such an additional example is shown. This piece was annealed for 1 hour at 500°C. and then compressed to 1.76%. It shows a considerable spreading of the plastic flow around the hole. Figs. 9-11 referring to hard aluminum might now be compared with Fig. 8.

Table 1. Various Percentages of Pre-compression for Aluminum Cylinders.

Cylinder No.	Percent Compressed	Final Load (lbs.)
10	0	0
9	3.0	15,000
8	5.0	18,500
7	10.0	26,200
6	15.0	31,500
5	20.0	34,500
2	26.5	.....
3	37.5	60,500
1	43.0	.....
4	43.8	66,500

Fig. 9 shows the result of compressing test piece No. 5 to 0.61% deformation. Also Fig. 10 shows a similar result of compressing test piece No. 6 to 0.99% deformation. Even for this small amount of deformation in comparison to 1.76% in Fig. 8, it can be seen in both cases that the flow layers extended from the hole to the edge of the test piece. This was accompanied by almost no plastic deformation between the layers. The polished surface had an appearance similar to that of mild steel in which two separate flow layers form without plastic deformation between the layers. However, on compressing the test pieces further, the entire triangular section between the layers became plastic as shown in Fig. 11.

IV. The Variation in Unit Shear Strain Along a Flow Layer of Mild Steel

Another question which was investigated in this connection is a determination of the magnitude of unit shear strain along a flow layer. Since flow layers of a very sharply defined character may be produced in mild steel just after the yield point is passed, this material was used for the purpose instead of aluminum. A rectangular compression test piece of mild steel containing a net of horizontal and vertical lines on the large flat surface as shown in Fig. 12 was then machined. The distances between the sides of the small squares were measured before testing on a comparator along sections inclined at 45° to the bar axis at three different points along the subsequent flow layer. The test piece was then stressed slightly beyond the upper yield point to about 0.147% compression on the basis of the test piece height as gage length. If all of the deformation is assumed to be confined to the flow layer itself, the average vertical percent compression would amount to about 1.65%. The distances were again measured between the sides of the same squares after deformation. The percent elongations e_x and e_y in the x and y directions (x—horizontal, y—vertical) and γxy the unit shear angle were then computed for three points along the flow layer. Using the following relation, the maximum unit shear along the layer was computed,

$$\gamma^2_{max} = (e_x - e_y)^2 + \gamma^2 xy$$

where γ_{max} is the maximum unit shear strain. This unit shear is shown plotted in Fig. 13, along the flow layer. The shear strain is a maximum near the hole and decreases toward the edge of the test piece.

V. Conclusions

From the tests herein described, it appears that the more cold deformation within certain limits an originally soft and ductile metal contains, the more pronounced will be the flow

layers if it is again stressed to the plastic limit. There does not seem to be very much change in the appearance of the flow layers above about 25% pre-compression.

It is also shown that by properly designing the test piece, it is possible to produce flow layers in work-hardened aluminum similar to those obtained in mild steel without much plastic deformation in the remaining portion of the test pieces.

The shear strain along a flow layer in a compression test piece of mild steel was shown to be a maximum near the hole in the test piece.

Acknowledgment

The author is greatly indebted to Dr. A. Nadai for his suggestion of this investigation as well as for his aid during its progress. He is also indebted to Mr. R. E. Peterson, Manager of the Mechanics Division of the Westinghouse Research Laboratories, for his encouragement of this kind of investigation.

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READERS' COMMENT

Dear Mr. Gillett:

There is an interesting point arising from your remarks on page 280 of the December, 1932 issue of METALS & ALLOYS on which you may like to have some comment. I refer to the discrepancy between the fatigue tests made on direct stress and bending machines and their relation to the results of tests made in air and in vacuo. Firstly, with regard to the tests in air, the discrepancy between the Haigh value and the Wöhler value is due of course to the redistribution of stress which takes place in the cantilever test, giving an apparently higher value in bending. With regard to the tests in vacuo, we have only as yet made these under direct stress conditions, and in the case of several materials a marked improvement occurs in vacuo. It is doubtful whether we shall repeat the results using bending stresses as considerable experimental difficulties are incurred but I have no doubt that if such tests were carried out, an improved performance would be obtained in vacuo. I mention this point because one not intimately acquainted with fatigue might infer that the improvement in vacuo is only related to direct stress tests, whereas it is extremely improbable that this is the case.

Yours sincerely,

H. GOUGH,
Superintendent

Jan. 13, 1933,
National Physical Laboratory
Engineering Department,
Teddington.

EDITOR'S NOTE

According to the experience of the Bureau of Standards, reported by France,[1] a uniform, ductile material would be expected to show only very small differences in endurance in axial loading and rotating beam tests. All too few comparisons have been made on copper base alloys. Irwin, as pointed out on page 201 of our September, 1932, issue, found slightly higher endurance on annealed copper in axial loading. If redistribution of stress in rotary bending is the source of the 20% discrepancy in Dr. Gough's axial and rotary beam tests of the same lot of annealed 70:30 brass in air, and not some idiosyncrasy of this particular lot, one would have expected the same effect to show up in Irwin's tests. Until the situation in regard to Haigh versus rotary testing of copper base alloys in air is made more clear, the writer feels that the enunciation of a new law of fatigue in vacuo, based on the behavior of one lot of material that appears to act peculiarly in air, should be made in mild and tentative terms, and its acceptance in toto should be deferred till further evidence is published.

While there would be great difficulty in testing a rotating beam specimen in vacuo, the method used by Dr. Gough on the Haigh machine might perhaps be applied to the Illinois spring machine,[2] in which the specimen is not rotated but is held rigidly at one end while the other is given a gyratory motion. It is to be hoped that Dr. Gough will extend his work to other types of loading and to materials of a type that his pioneer work would indicate might be affected by atmospheric oxidation, but which show a more normal relationship in axial and rotary loading.

Though Dr. Gough states that "in several materials a marked improvement occurs in vacuo," the writer would appraise the differences cited by Gough and Sopwith in their 1932 Institute of Metals paper as so minute as to be meaningless save in the single case of the Bl Fl 70:30 brass. Perhaps further work has been done but not yet published.

It would not be the only time that the writer has been conservative in accepting, and outspoken in asking for further proof of, an important pronouncement in the endurance field, for he remembers taking that same attitude[3] when Dr. McAdam presented his early work on corrosion fatigue. McAdam and many others have since produced a great mass of information that has vastly advanced the knowledge of fatigue and thoroughly justified his first statements. It is hoped that Dr. Gough will even out-McAdam McAdam in adding to further knowledge.

As an individual, the writer is deeply interested in Dr. Gough's experiment, but when it comes to picturing the endurance situation in a Correlated Abstract, he feels that a conservative view must be taken.

H. W. GILLETT

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Editor, METALS & ALLOYS:

Concerning your editorial comment of the January issue, "Supremacy Areas," in the main the gist is just good sense applied to marketing and distribution. Tho here and there attempts are made to make one article, material or service supplant another needlessly, few result-expectant sales managers would persist in chasing butterflies.

However, in the third paragraph of your clarification of the idea, in the one quick statement, "all that is necessary is to state the facts clearly so that they will be understood and accepted," is it not possible that one essential is being overlooked, that is to whom and where and when?

We submit that today's problems are not so much the lack of science in the ability to produce but in the ability to search out the intermediate or ultimate consumer who is susceptible to the supremacy of the product in question.

"The better mouse trap" idea still holds, but, millions of them might be made and pile up and rot if some means of finding the needy one and reaching his consciousness were not used.

The idea of supremacy areas offers possibilities for fair profit to the manufacturer and a fair exchange to the user, but it also suggests the need for market research into the user's province, after which a not unimportant factor would be the development of an instrument to disseminate the facts, whether that instrument be personal approach, printed media, existing distributing organizations and their personnel or what not. In this latter many fail.

Finally, supremacy might be "real," or it might be conceded by technical experts, and yet it might suffice the producer nothing. On the other hand, in that enlightenment spreads in peculiar ways, the supremacy might be vested in popular "acceptance," and from a profit standpoint not to be ignored, even tho steps were being taken to earn supremacy in a rightful field.

Yours very truly,

LEWIS HERMAN.

Herman and Company,
Youngstown, Ohio.
Jan. 23, 1933.

EDITOR'S NOTE

Far be it from us to deny the importance of bringing "messages of supremacy" to the attention of those who should utilize the virtues of the material. We heartily agree with Mr. Herman's point concerning "to whom and where and when." Probably we should have written "to state the facts clearly *and* so that they will be understood and accepted."

As to "ability to search out the consumer," we believe that it is displayed by such advertisements as those of International Nickel, New Jersey Zinc and General Electric in our December, 1932, issue. Vital engineering facts are there presented to the readers of METALS & ALLOYS who constitute a group of engineers and executives concerned with finding the right materials for specific applications. Such messages are more effective in influencing our readers than vague glittering generalities.

RICHARD RIMBACH

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[1] R. D. France. Endurance tests of steel, comparison of results obtained with rotating beam versus axially loaded specimens. *Proceedings, American Society for Testing Materials,* Vol. 31, Part 2, 1931, pages 176-183.

[2] H. F. Moore. The fatigue of metals. McGraw-Hill Book Co., 1927, page 99.

[3] Discussion *Proceedings American Society for Testing Materials,* Vol. 27, Part 2, 1927, pages 141-143.

Single Crystals of the Intermediate Compound Antimony-Tin

H. S. Van Klooster

By H. S. VAN KLOOSTER* and M. O. DEBACHER*

M. O. Debacher

ALTHOUGH a number of metals have been obtained in the form of single crystals, no attempts have been made except incidentally to prepare single crystals of intermetallic compounds. The only reference in the literature was an article by Sizoo and Zwicker[1] who succeeded in obtaining alloys of nickel-iron 1 mm. thick and about 20 cm. long, consisting entirely or partly of a single crystal. These workers let the molten alloy rise gradually into a quartz tube during freezing. This appears to be a modification of the method originally introduced by Czochralski[2] which consists in pulling a glass rod, in contact with the molten metal, at a uniform velocity upward at such a rate that the metal does not break. This procedure was tried out in the case of the compound Sb-Sn but the results were not entirely satisfactory. The method which was finally adopted was the one proposed by Obreimow and Schubnikow[3] in which a two-compartment vertical tube furnace with two separate heating coils is employed. The compartments are separated by a copper plate with an opening in the center. The temperature in the furnace was measured by means of thermocouples placed near the top and bottom of the upper compartment and in the center of the lower compartment. A pyrex tube of about 3 mm. inside diameter with a capillary tip was used to hold the alloy.

The sample tube was filled with a stoichiometric mixture of Sn and Sb. A little charcoal to prevent oxidation was added to the mixture. The tube was next introduced into the furnace in such a way that the capillary end protruded through the hole in the copper plate and extended from 1 to 2 mm. into the bottom compartment. The temperature of the furnace was raised to 480°C., this being 50° above the temperature at which the compound is formed. The bottom of the lower compartment was removed and the temperature slowly decreased so as to start crystallization in the tip of the tube. The rate of cooling from 480° to 300°C. was diminished in successive experiments until finally after increasing the period of cooling through this interval from ½ hr. to 15 hrs. a single crystal of about 5 cm. length was obtained. The temperatures at the top and at the bottom of the upper compartment, and that at the center of the lower compartment during the 15 hour run are recorded in Fig. 1. In order to get this exceedingly long period of crystallization it is necessary to keep both compartments closed and lower the temperature in the top section very gradually as is indicated by curve 1 (Fig. 1). The compound, being exceedingly brittle, was freed from the glass tube with hydrofluoric acid. The resulting rods were mounted lengthwise in sulphur molds, ground and polished, and then etched with a solution of nitric and acetic acids.

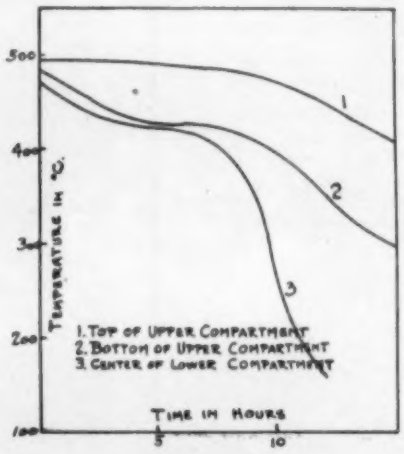

Fig. 1

The effect of different cooling times is strikingly demonstrated in Figs. 2 to 6. Fig. 7 shows side by side a piece of a polycrystalline aggregate and a cleavage plane (in the direction of growth) of a single crystal. From the measurement of the angle of contact between neighboring cleavage planes which was in all cases about 60°, it was concluded that the compound possibly belonged to the cubical system with cleavage along rhombo-dodecahedral planes, the angle between (110) and (101) being exactly 60°. To obtain further information concerning the structure some of the crystals were powdered to a fineness of 200 mesh and an X-ray spectrogram made using a

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Fig. 2. Cooling time ½ hr.

Fig. 3. Cooling time 2 hrs.

Fig. 4. Cooling time 3 hrs.

Fig. 5. Cooling time 6 hrs.

Fig. 6. Cooling time 15 hrs.

Fig. 7. Comparison of the fracture of multiple and single crystals.

quadrant cassette†. The pattern obtained showed that the compound Sb-Sn possesses a simple cubic structure which is not unusual for salts of the NaCl type, but extremely rare in the field of intermetallic compounds. The value for the atomic distance in a unit cell with 4 Sb and 4 Sn atoms at the corners of the cube is 3.061 A.U. The density computed on the basis of this value (assuming $N=6.06\times10^{23}$) is 6.918. The value found by experiment (at 20°C.) was 6.9104 ± 0.005. While the work was in progress a brief note by Atomi Osawa[4] brought out the fact that Sb-Sn alloys over the range 43-55% consist of a solid solution of one of the metals in the compound Sb-Sn. The atomic distance for the 50% Sb-alloy was given as 3.061 A.U. ($\pm$ 0.007) and the density as 6.9109 in good agreement with the values found above. A later X-ray investigation by Bowen and Jones[5] which covers the whole binary system Sb-Sn gives the value for the atomic distance in the compound Sb-Sn as being 3.060 A.U. ($\pm$ 0.03) and the lattice constant (on the basis of 4 Sb and 4 Sn atoms per elementary cube) as 6.120 A.U. The last named workers quote an earlier determination of the lattice constant by Goldschmidt[6] who found the value 6.130 A.U. ($\pm$ 0.007). It is interesting to note that the atomic distance found in the present and in the preceding investigations lies midway between the values of atomic distances reported by Goldschmidt[7] for Sn and Sb atoms (2.79 and 3.37 A.U. respectively).

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[1] *Zeitschrift für Metallkunde*, Vol. 3, 1929, page 125; *Metals & Alloys*, Vol. 1, 1929, page 29.
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[3] *Zeitschrift für Physik*, Vol. 25, 1924, page 31.
[4] *Nature*, Vol. 124, 1929, page 4.
[5] *Nature*, Vol. 126, 1930, page 846; *London, Edinburgh & Dublin Philosophical Magazine & Journal of Science*, Vol. 12, Series 7, 1931, page 441.
[6] *Berichte der deutschen chemischen Gesellschaft*, Vol. 60, 1927, page 1263.
[7] *Berichte der deutschen chemischen Gesellschaft*, Vol. 60, 1927, pages 1289-1290.

†For this part of the investigation the authors are indebted to Prof. G. L. Clark of the University of Illinois under whose direction the X-ray work was carried out.

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C. B. Templeton, formerly Assistant to the President, is now Assistant to the Vice-President in Charge of Sales, Ludlum Steel Company, Watervliet, New York. Mr. Templeton has been associated with the Ludlum Steel Company in various capacities since his graduation from Yale in 1920. Prior to his last post as Assistant to the President, he served for several years as Assistant Manager of Sales.

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O. B. Wilson, formerly manager of the Cleveland, Ohio office of The Brown Instrument Company, has been appointed District Manager of the territory comprising Texas, Louisiana and Southern Arkansas, with headquarters at Houston, Texas.

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Merrill G. Baker, recently Assistant to the Vice-President of Ford, Bacon & Davis, Inc., has been elected President of the Union Steel Casting Company of Pittsburgh.

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